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HP9-4-.30 WELD PROPERTIES AND MICROSTRUCTURE

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HP9-4-.30, ultra high-strength steel, the case material for the Advanced Solid Rocket Motor (ASRM), must exhibit acceptable strength, ductility, toughness, and stress corrosion cracking (SCC) resistance after welding and a local post weld heat treatment (PWHT). Testing, to date, shows that the base metal (BM) properties are more than adequate for the anticipated launch loads. Tensile tests of test specimens taken transverse to the weld show that the weld metal overmatches the BM even in the PWHT condition. However, there is still some question about the toughness and SCC resistance of the weld metal in the as welded and post weld heat treated condition.

To help clarify the as welded and post weld heat treated mechanical behavior of the alloy, subsize tensile specimens from the BM, the fusion zone (FZ) with and without PWHT, and the heat affected zone (HAZ) with and without PWHT were tested to failure and the fracture surfaces subsequently examined with a scanning electron microscope (SEM). Table I shows the test results (the average of 5-6 test specimens) and, for comparison, average results from a large number of full scale BM tensile tests accomplished at MSFC. The full size specimen tensile test results and the subsize specimen tensile results for the BM are comparable indicating at most a small specimen size effect for the subsize tensile results. Without PWHT the FZ and HAZ materials have a reduced 0.2% offset yield strength (YS), a very high ultimate tensile strength (UTS), and significant loss of ductility as measured by either %El or %RA. The reduced YS is probably due to the presence of appreciable amounts of relatively soft retained austenite that begins to plastically deform at a lower stress than the higher strength martensite matrix surrounding it. The influence of retained austenite on the fracture of maraging steels

TABLE I

		Subsize (0.16" dia.)				
Property	NASA full size	BM	FZ,NoPWHT	FZ,PWHT	HAZ,NoPWHT	HAZ,PWHT
YS(ksi)	206	211	181	221	188	215
UTS(ksi)	227	237	271	250	293	251
%El (1/2" GL)	22.8*	14.3	7.5	14.6	8.2	16.2
%RA	54.3	54.3	35.5	43.6	45.2	57.2

* The gage length (GL) for the NASA full size specimens was 2".

has been reported in the literature.[1] It was suggested that the highly strained pools of austenite contain carbide or other particles that are sites for the creation of voids which grow, coalesce, and result in failure by microvoid coalescence. The data in Table I, also, clearly show that the PWHT used (950 °F for 2 hours) results in return of the FZ and HAZ to almost base metal YS and UTS values. The %El and %RA of the HAZ also indicate recovery of most of the BM ductility, but the %RA value for the post weld heat treated FZ are considerably below the BM values indicating the ductility of the FZ remains low. These results would suggest that a good PWHT should return the HAZ to desired strength and toughness, but probably not recover the FZ toughness.

SEM examination of the fracture surfaces of representative samples from the subsize tensile tests (BM, FZ with and without PWHT, and HAZ with and without PWHT) shows the following results and trends. All fracture surfaces exhibit a microvoid coalescence failure mode. In general, there appears to be a bi-modal distribution of dimple or void sizes on the fracture surface with small dimples being less than 1 micron in diameter and the large dimples being from 1-5 microns in diameter. Another clearly discernible characteristic was that the FZ and HAZ without PWHT specimens had a much higher fraction of the small dimples on the fracture surface than the BM. After PWHT, the HAZ showed an increase in the fraction of larger dimples while the FZ seemed to show a decrease. This reduction in FZ dimple size is probably due to sample to sample variation and should not be interpreted as caused by the PWHT. Preliminary estimates of the average dimple size on the fracture surface (D_0), which depends on the dimple size distribution, and the strain to fracture estimated from %RA (e_f) are given in Table II. These data show that as D_0 increases, at least within the range of sizes considered here, e_f increases. Garrison, et al,[2] have related void or dimple spacing to crack tip opening displacement. As dimple spacing increases dimple diameter will increase in direct proportion and, since, crack tip opening displacement is proportional to strain to fracture the data in Table II should follow a relationship similar to that determined by Garrison. He showed, at low void spacings, the relationship between void spacing and crack tip opening displacement should be linear; experimental data appeared to confirm this result. Thus,

TABLE II

Material	D^o (micron)	e_f
BM	1.7	0.78
HAZ no PWHT	1.3	0.42
FZ no PWHT(1)	1.1	0.44
FZ no PWHT(2)	1.2	0.47
HAZ PWHT	1.4	0.70
FZ PWHT	1.0	.49

for the dimple sizes in this work, it is expected that the relationship between e_f and D_0 should be linear with a zero intercept. The data do indicate a linear least squares fit of $e_f=0.4D_0$. It appears from these data that the ductility of the HAZ is regained to a large extent by the PWHT while that of the FZ is only slightly improved. Consequently, the weak link, as far as toughness and stress corrosion cracking of the post weld heat treated weldment is concerned, would seem to be the fusion zone.

Limited tensile SCC tests using the ASTM 3.5% NaCl Alternate Immersion and the 5% Salt Spray Tests were performed at MSFC and indicate a possible problem in the weld metal. In the as welded condition most of the failures occurred in three weeks or less at stresses equal to or greater than 75% YS with the cracking tending to initiate in the HAZ. After PWHT there appeared to be an improvement, but failures still occurred at close to three weeks in some cases. However, after the PWHT the failures seemed to initiate more in the FZ. Several specimens were sectioned perpendicular to the fracture surface, polished, and etched in order to determine the location of the fracture initiation. These specimens were also used for microhardness measurements to determine if there was a relationship between hardness and the location of the crack initiation site. Gouch[3] has concluded that, for high strength steels, welding greatly increases susceptibility to SCC with the hardest regions of the weld zone being the most susceptible with microstructure having a secondary effect. He also found that PWHT improves SCC resistance since it results in a reduction in hardness in the FZ and HAZ.

Comparison of the microhardness data, taken in this instance from a limited number of the HP9-4-.30 weldments tested in SCC, with the apparent crack initiation point indicates that in the as welded condition the crack does appear to initiate in one of the harder regions of the FZ or HAZ. However, the PWHT seems to shift the failure location into the FZ even though there are still harder regions in the HAZ. Apparently, in the post weld heat treated condition the microstructure and/or microsegregation become more dominant than the hardness in controlling the SCC. Of course, the effect of the surface condition may also play a major role, and it may be that in the PWHT condition the exposed surface of the FZ is more susceptible to pitting so it is easier to initiate the cracks. Once the pits (or other surface defect) has been initiated the crack probably grows by a hydrogen embrittlement mechanism. Tromans[4] studied the stress corrosion cracking of HY-180 steel at various corrosion electrochemical potentials and found in all cases that the crack propagation was consistent with hydrogen embrittlement.

It has been suggested that a higher temperature PWHT and perhaps a different temper during BM processing could improve the SCC resistance of the weldments. It is possible to achieve some improvement, but if the FZ susceptibility problem is associated with the microstructure/microsegregation then that cannot be changed by a PWHT.

REFERENCES

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